UNCLASSIFIED

AD NUMBER
ADB282381
NEW LIMITATION CHANGE
TO Approved for public release, distribution unlimited
FROM Distribution: Further dissemination only as directed by AFOSR/NA, 801 N. Randolph St., Rm. 732, Arlington, VA 22203-1977, Jul 2002; or higher DoD authority.
AUTHORITY
Corrected SF 298, 8 Jul 2002

THIS PAGE IS UNCLASSIFIED

REPORT DOCUMENTATION PAGE

AFRL-SR-AR-TR-02-

0322

1. REPORT DATE *(DD-MM-YYYY)* 08/07/2002	**2. REPORT TYPE** Final Report	**3. DATES COVERED** *(From - To)* 10/15/98 - 4/14/02
4. TITLE AND SUBTITLE "Mechanics Models for Strength and Durability of Fiber-Reinforced Ceramic and Polymer Composites"		**5a. CONTRACT NUMBER**
		5b. GRANT NUMBER F49620-99-1-0027
		5c. PROGRAM ELEMENT NUMBER
6. AUTHOR(S) William Curtin		**5d. PROJECT NUMBER**
		5e. TASK NUMBER 2302/BX
		5f. WORK UNIT NUMBER
7. PERFORMING ORGANIZATION NAME(S) AND ADDRESS(ES) Brown University, Division of Engineering, 184 Hope Street - Box D, Providence, RI 02912		**8. PERFORMING ORGANIZATION REPORT NUMBER**
9. SPONSORING / MONITORING AGENCY NAME(S) AND ADDRESS(ES) AFOSR/NA, 801 N. Randolph St. Rm 732, Arlington, VA 22203-1977		**10. SPONSOR/MONITOR'S ACRONYM(S)**
		11. SPONSOR/MONITOR'S REPORT NUMBER(S)

12. DISTRIBUTION / AVAILABILITY STATEMENT

Approved for public release;
distribution unlimited.

20020815 095

13. SUPPLEMENTARY NOTES

14. ABSTRACT

New advances in the field of multiscale modeling of the tensile strength and time-dependent failure behavior of fiber reinforced composites are described. The major accomplishments include the development of a methodology for extracting information from finite element studies at the scale of a few fibers and using this information in a flexible Green's function model to compute damage evolution and failure in the composite. Recent work as aimed at developing models to predict composite electrical resistance as a function of mechanical damage so that electrical resistance measurement of in-service components can be used as an on-board health-monitoring system.

15. SUBJECT TERMS

16. SECURITY CLASSIFICATION OF:			**17. LIMITATION OF ABSTRACT**	**18. NUMBER OF PAGES**	**19a. NAME OF RESPONSIBLE PERSON** William Curtin
a. REPORT	**b. ABSTRACT**	**c. THIS PAGE**		20	**19b. TELEPHONE NUMBER** *(include area code)* 401.863.1418

Standard Form 298 (Rev. 8-98)
Prescribed by ANSI Std. Z39.18

Mechanics Models for Strength and Durability of Fiber-reinforced Ceramic and Polymer Composites

Air Force Office of Scientific Research
Grant # F49620-99-1-0027

FINAL REPORT
Progress Report for the Period
October 15, 1998 – April 14, 2002

W. A. Curtin
Division of Engineering, Brown University
Providence, RI 02912
Tel: 401-863-1418 FAX: 401-863-9025
e-mail: curtin@engin.brown.edu

ABSTRACT

New advances in the field of multiscale modeling of the tensile strength and time-dependent failure behavior of fiber reinforced composites are described. The major accomplishments include the development of a methodology for extracting information from finite element studies at the scale of a few fibers and using this information in a flexible Green's function model to compute damage evolution and failure in the composite. Recent work as aimed at developing models to predict composite electrical resistance as a function of mechanical damage so that electrical resistance measurement of in-service components can be used as an on-board health-monitoring system.

1. Objectives at Outset of Grant Period

Develop coupled micromechanical, mesomechanical, and macromechanical models of strength and lifetimes of fiber-reinforced composites.

Micromechanical scale: use three dimensional finite element studies to predict the stress redistribution around fiber breaks as a function of stress, fiber/matrix interface conditions, matrix deformation, and time.

Mesoscale: utilize numerical simulation techniques and fiber strength statistics to predict (i) the damage evolution at the fiber-to-fiber scale and (ii) the ultimate composite strength as a function of the micromechanical load transfer found at the microscale, the fiber geometry, fiber strength statistics, and time-dependent properties.

Macromechanical scale: develop size-scaling relations for composite strength that are accurate to very large sizes, and construct macroscopic models of lifetime using remaining strength as a measure of damage. Incorporate mesoscale models into macromechanical structural models to study damage tolerance under realistic loading conditions.

2. Accomplishments during Grant Period

Tremendous progress was made in achieving our objectives, with significant accomplishments in the development of realistic, predictive, multiscale models of failure in fiber-reinforced composites, including static and time-dependent failure, and encompassing all three major composite material systems: ceramic matrix composites (CMCs), polymer matrix composites (PMCs), and metal matrix composites (MMCs). This cumulative body of work represents a significant contribution to the field of composite mechanics. As a result, the PI does not believe further research in the area of fiber-dominated failure properties is necessary or warranted at this time. Future work should emphasize further applications of these results and methods to practical materials systems and use of this work as a basis for developing insight into damage detection and system life-prediction prognosis of structural systems.

Below, the major accomplishments of this effort are summarized as follows. Basic ideas and models are described and applied to polymer matrix composites first; some of the relevant models were developed during the previous granting period and are not discussed in detail. Multiscale modeling efforts to more precisely bridge the micro, meso, and macro scales of damage evolution are then described, with applications to titanium matrix composites. Work related specifically to damage evolution and degradation in ceramic matrix composites is then presented. Finally, our most recent work on modeling damage detection in PMCs by electrical resistance methods is described. This work represents the preliminary efforts taking the mechanical models developed previously and their associated insights into important areas associated with component evaluation and a priori life prediction. All of the work reported below has been published in the technical literature, and hence many details and results are omitted from this report. A list of publications is provided in Section 3; a list of invited presentations is given in Section 4; and a list of personnel supported on the grant is given in Section 5.

2a. ***Influence of Composite Dimensionality on Failure Strength***

3d simulation models demonstrate that there are notable differences between the failure of 2-d tapes (single ply specimens) and "bulk" 3d specimens, due primarily to the difference in stress concentration factors in 2d and 3d. Figure 1 shows the strength distributions for the 2d 1-ply specimens of width 40 or 160 fibers, for 4-ply specimens of width 40 fibers, and for a "bulk" 3d system of 160 fibers. The 2d material is weaker and less reliable than the 4-ply material, which approaches the "bulk" strength distribution. The strengths of much larger composites can be determined by appropriate size-scaling. The broader strength distributions of the 2d systems leads to even larger discrepancies in strength between 2d and 3d composites with increasing composite size. **These results have important implications for the modeling of composite lifetimes under conditions of fiber strength degradation. Specifically, since fiber degradation is often a highly non-linear function of the local stress state, attempts to model real composites using 2d models are expected to lead to enhanced degradation rates and lifetimes that are much shorter than, and scale differently than, lifetimes obtained on more-realistic 3d composites.**

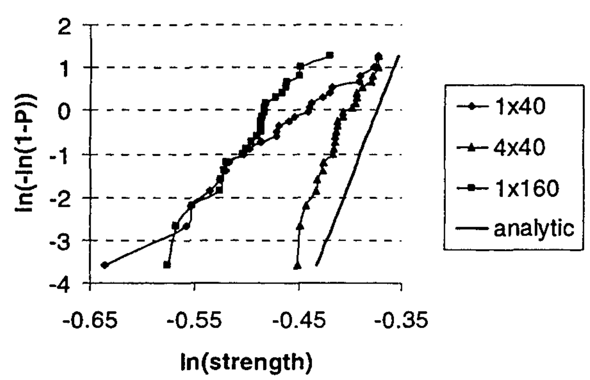


Figure 1. Weibull plot of the probability of failure vs. strength for 2- and 3-dimensional composites. 1x40 and 1x320 are 1-ply 2-d tapes; 4x40 is a 4-ply tape; analytic line is for "bulk" 3d system with 160 fibers. Fibers have Weibull modulus = 5.

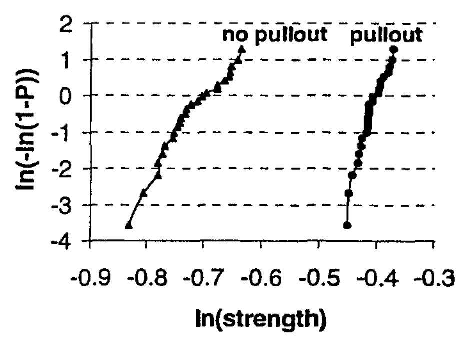


Figure 2. Weibull probability of failure vs. strength for 4x40 4-ply tape with, and without, fiber pullout. Fibers have Weibull modulus =5.

The fiber "pullout" stresses, i.e. the finite stresses carried by a broken fiber within the slip zone along the broken fiber, play an important role in composite strength as well. Classic models such as the "chain-of-bundles" models neglect this load-carrying capability. Simulations of various composite geometries with and without these pullout stress show how such simplifying assumptions affect strength predictions. Figure 2 shows the strength distributions for 4-ply composites with 40 fibers per ply: without fiber pullout stresses, the strength of the composite is reduced substantially and the reliability is reduced. The latter leads to even larger differences in composite strength with increasing composite size. **Hence, the proper inclusion of residual sliding or other forces along broken fibers and the three-dimensional distribution of fiber breaking**

in the composite are quite important in obtaining quantitatively accurate composite strengths.

2b. Tensile strength of Carbon-reinforced polymer matrix composites

A new model to describe the non-Weibull strength scaling typical of carbon fibers. Such basic models are key input to making accurate predictions of composite strength. The model assumes that each individual fiber obeys a standard Weibull model with Weibull modulus ρ' and fiber-specific characteristic strength σ_o^i. The fiber-to-fiber variations in σ_o^i are then assumed to follow a second Weibull distribution with Weibull modulus m and characteristic strength $\overline{\sigma}_o$. The resulting "Weibull of Weibulls" distribution for a collection of fibers exhibits a Weibull distribution at fixed gauge length with a Weibull modulus of $\rho \approx m\rho' / (m^2 + \rho'^2)^{1/2}$ but a length scaling of the characteristic strength that is controlled by the value of ρ', as found in carbon and other fiber materials. **This development provides a physical interpretation of the non-standard strength statistics and also permits composites composed of such fibers to be studied via numerical simulations, which is something that was not possible using previous empirical relationships.**

The consequences of this modified fiber strength scaling on the overall composite failure have been investigated numerically and analytically. Analysis of composite failure under conditions of "Global Load Sharing", wherein failed fibers transfer their loads equally to all surviving fibers, can be performed analytically and demonstrates the basic decline in strength due to the correlation of fiber strengths along the fiber length. Numerical simulations using our "Local Load Sharing" codes confirm this qualitative picture and provide quantitative strength values. Moreover, from our previous work on analytic strength models for standard Weibull fibers, we have formulated a fully analytic model to predict the strength distribution and size-scaling of strength of fibers composed of the present "Weibull of Weibull" fibers. Figure 3 compares our fully analytic model to numerical simulation data for several fiber strenghth parameter values, and the agreement is seen to be excellent.

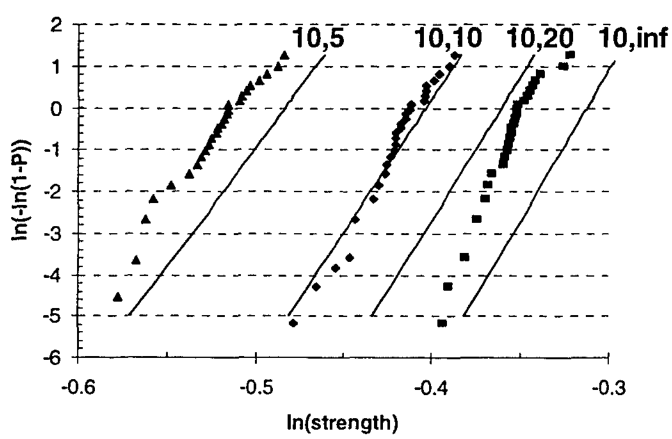


Figure 3. Strength distributions, in Weibull form, for composites composed of Weibull of Weibulls fibers, with Weibull parameters ρ' and m as indicated. Solid lines are analytic predictions; symbols are simulation results.

Based on the above success, the analytic model has been applied to predict the composite strength of several graphite/epoxy systems. The fiber strength properties and appropriate in-situ gauge length have been obtained from literature data on the fibers and on single fiber composites. Table I shows the fiber strength data and the composite strength predictions and measured values. The quantitative agreement is outstanding.

	Fiber Strength	Ineff Length				Vol. Frac.	Fiber Mod.	σ_{uts}	σ_{uts}
System	σ_c	δ_c	ρ'	m	ρ	f	E_c	Expt	Theory
AS-4/ 828mPDA	5783	501	10.7	8	6.4	0.59	138	1890	1990
AS-4/ 828mPDA	6437	501	8.33	6.3	5.0	0.59	138	1890	2139
AS-4C/ 828mPDA	5863	458	10.7	8	6.4	0.64	150	2044	2185
T300/Epicote	6016 (5346)	378	7	7	5	1	232	2880	3529 (3189)
T300/ #3601	5734 (5181)	378	7	7	5	0.59	138	1575	1885 (1703)
T300/ #3631	6016 (5436)	378	7	7	5	0.59	136	1740	1978 (1787)

Table I. Constitutive and composite data for various systems, and measured and predicted tensile strengths. Data in parenthesis is obtained using estimated thermal stress corrections in the s.f.c. test. Units are as follows: σ_c(MPa), δ_c(μm), E_c(GPa), σ_{uts}(MPa).

2c. Applicability of Shear Lag Models for Fiber Composites

Shear-lag models have been widely used to model fiber composite failure due to their computational efficiency and the fact that the models satisfy equilibrium and compatibility. However, in various situations, the shear-lag models are not accurate. With regard to composite deformation and failure, a critical aspect of any model is the stress transfer from broken to unbroken fibers. Thus, the stress transfer in fiber-reinforced polymer-matrix (PMC) and aluminum-matrix (AMC) composites was studied using a detailed 3d finite element model (FEM) and using the standard shear-lag model (SLM) to determine under what situations the SLM can and cannot be used with confidence. The predictions of the SLM for stress distributions along a broken fiber were found to agree well with the FEM results for elastic/plastic matrices. The stress transfer predicted by the SLM is in good agreement with the FEM results for the PMC at high fiber volume fractions but not at low volume fractions. For the AMC, the SLM performs poorly in general: at high fiber volume fractions, the near-neighbor stress transfer is much lower than found by FEM and the stress transfer to the farther neighbors is larger than found in by FEM; at low fiber volume fractions, cancellation of various effects leads to spurious agreement between the SLM and FEM. These results suggest that the SLM is accurate only for high fiber/matrix stiffness ratios and high fiber volume fractions. The neglect of (i) the finite fiber dimensions and fiber shear deformation and (ii) matrix load carrying capacity are shown to be responsible for the deviations in the shear-lag model. **It is concluded that the standard SLM cannot generally be used for simulations of**

composite deformation and failure, although it is accurate for a limited range of composite constituent material properties.

2d. Accuracy and Efficiency of Green's Function models for Composite Failure

Our previously-developed Green's function method (GFM) for simulating fiber damage evolution and tensile strength in fiber-reinforced composites was compared in detail to the predictions of the standard shear-lag model widely used in the literature. Since the shear-lag model has limited applications while the GFM is more general, but possibly less accurate, it is valuable to assess the GFM. The GFM extracts the stress concentration factors describing how broken fibers redistribute load to surrounding unbroken fibers from more-detailed micromechanical models, such as finite-element models, and uses this information to then calculate the propagation of fiber damage up to composite failure. The GFM only approximately includes the three-dimensional nature of the stress transfer and uses an approximate superposition method, but reduces the computational problem significantly. In this work, an elastic 3-d shear-lag model was used to provide the input to the GFM, and then the predictions of fiber damage and failure from the GFM were compared directly to those from full 3-d simulations using the complete shear-lag model. For exactly the same starting configuration of stochastic fibers, the GFM predicts (i) evolution of the fiber damage and the formation of critical clusters that are nearly identical to those found in the shear-lag model (Figure 4), (ii) composite tensile strengths within 2% of the shear-lag model, (iii) a Weibull modulus for the composite strength that is essentially the same as found for the shear-lag model, all while requiring over an order of magnitude less computational time for modest-size composites. The accuracy of the GFM was shown to be even better for a more-realistic non-elastic problem involving interface/matrix shear-yielding. **The important conclusion from this work is that the GFM is well-suited for tackling a wide range of problems that *cannot* easily be studied using the shear-lag model; e.g. bending deformation and failure, matrix crack propagation, fatigue crack growth, and other situations in which the fiber stress distribution is non-uniform even in the absence of any fiber damage.**

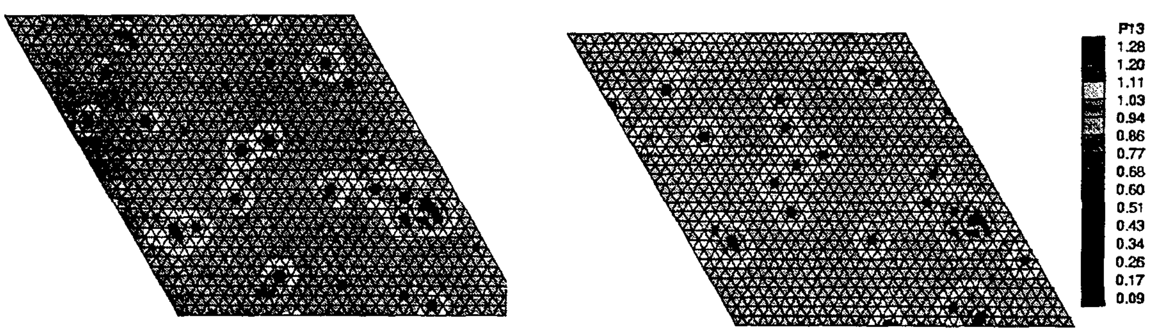


Figure 4. Stress distribution in the plane of critical fiber damage just at point of failure onset, as predicted by the shear-lag model and the Green's function model. Blue indicates broken fibers; Yellow-Red indicates high stress concentrations on surviving fibers.

2e. *Multiscale Modeling of Fiber Composite Failure*

We have developed a multiscale approach to composite failure in which detailed information on small-scale micromechanics is incorporated approximately yet accurately into larger-scale models that are capable of simulating extensive damage evolution and ultimate failure. Specifically, we employed a fully three-dimensional finite element model to investigate the load transfer from broken to unbroken fibers as a function of applied stress and interface friction coefficient (Figure 5). With a Von Mises matrix yield criterion, constraint effects permit the matrix to carry some of the transferred load near the fiber break, a feature not captured in previous shear-lag models. The single-break results for stress concentrations were then used as the discrete Green's Functions for load transfer in a Green's Function Model for the full composite, and the predicted load transfer around a seven-fiber-break cluster was shown in good agreement with additional finite-element results. The GFM was then used to predict overall damage evolution and composite failure for an IMI-834 Ti/SCS-6 SiC system. We investigated the role of the interface friction coefficients within this system, as well. However, for this system, the composite tensile strength is rather insensitive to the friction coefficient due to a cancellation of factors. For values of μ comparable to those measured experimentally, the predicted tensile strength was found to be in excellent agreement with the measured value *with no adjustable parameters* (Table 2). Analytic models for scaling of the tensile strength to very large sizes were then shown to agree well with strengths obtained from extensive simulations. **This work represents the first detailed hierarchical coupling approach to damage evolution and forms the basis for used attacking a wide variety of damage and failure problems in fiber composites.**

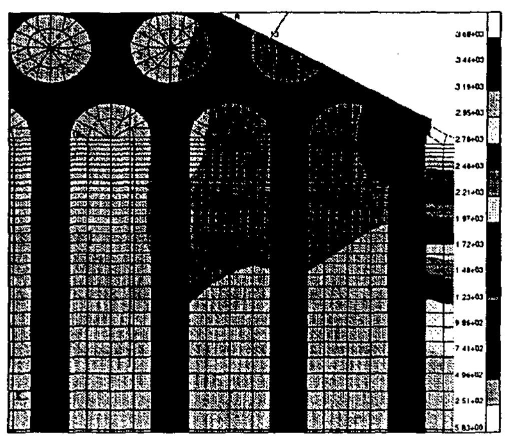


Figure 5. 3d stress distribution determined by FEM around a broken fiber in Ti-MMC

Parameters	*Prediction**			*Experiment*
Coefficient of friction, μ	0.25	0.5	0.9	-----
Interfacial shear stress, τ, MPa	51.7	103.5	186.3	55-75
Tensile strength, σ_{uts}, MPa	2268.0	2317.5	2344.5	2200-2300

Table 2. Predicted and measured Ti-MMC tensile strengths

2f. *Multiscale Modeling of Fatigue Failure in Fiber Composites*

Low-cycle fatigue failure in Titanium metal matrix composites is caused by two separate damage mechanisms: fatigue crack growth in the Ti matrix and fiber breakage. Following upon our success with a multiscale model for the tensile strength, here a related multiscale numerical model for predicting both crack growth and fiber breakage was developed and applied to predict low-cycle fatigue lives in a SiC-fiber reinforced Ti matrix composite. A 3-d finite element model containing a matrix crack bridged by SiC fibers was used to calculate both the matrix crack tip stress intensity factor and the local fiber stress concentrations due to the matrix crack as a function of the crack size. The crack tip stress intensity factor was used in a Paris-law model for the growth rate of the matrix crack. The matrix fatigue crack was assumed nucleated on the first loading cycle by the formation of a crack in the reaction layer around a fiber. The local stress distributions in the fibers were then used as the effective "applied" load within a 3d Greens Function method that simulates the fiber damage process at any fixed fatigue crack size. The appropriate Green's functions for fiber-to-fiber load transfer were also determined from the finite-element models. Methods to incorporate the additional stress concentrations on the fiber surfaces within the fatigue crack, as found in the CMC case in topic 1 above, were also developed. The results show that fiber failure occurs preferentially within the matrix crack region, where the fiber stresses are comparatively high. Composite failure occurs when the damage in this region is sufficient to drive fiber failure throughout the remainder of the composite in a crack-like fracture mode. A fatigue life threshold was predicted at about 80% of the quasistatic tensile strength. Below this stress, the fiber bundle can survive even with a matrix crack extending throughout the entire cross-section of the sample and fatigue failure is then driven by fatigue degradation of the fiber strengths. Predictions for the low-cycle fatigue of Ti-matrix (IMI834) reinforced with SCS-6 SiC fibers compare well with available experimental data at high stresses using pristine fiber strengths *and no adjustable parameters* (Figure 6).

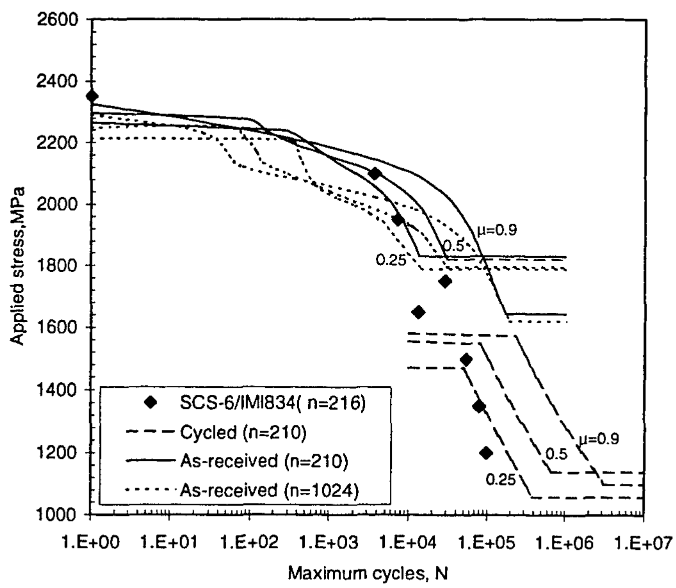


Figure 6. Predicted (lines) and measured (points) cyclic fatigue lives for Ti-MMC.

The experimental data are well matched at lower stresses by using literature values for the fatigued fiber strength *and no adjustable parameters*. The model also demonstrates that fatigue life can be dependent on the actual composite size and can be very sensitive to initial fiber damage.

This multiscale model makes use of several convenient features of the Green's function method. The method can deal with arbitrary local distributions of "applied stress" stemming from actual applied fields or damage not associated with fiber breakage. Although not investigated here in any detail, the statistical distribution of fatigue life and the role of processing damage to the matrix and/or fibers, as well as other factors, can be investigated in some detail using this multiscale model approach. **The method here has clear applications to a wide range of other damage problems that involve multiple length scales in composites.**

2g. Time-Dependent Failure of Composites under Local Load Sharing

The Green's function models for fiber damage evolution and composite failure for time-independent processes have been extended to permit time-dependent fiber degradation. We have developed codes to permit slow crack growth degradation in the fibers that is a function of the local stress acting on the flaws in the fiber as the composite damage evolves. Preliminary simulation results for the composite failure and lifetime show that, at low composite stresses, considerable fiber damage must occur to induce overall composite failure. A large fraction of the composite cross-section, for 900 fibers, must fail prior to a rapid acceleration of damage and total failure. Unfortunately, this makes the simulations very computationally-intensive, so that a complete statistical data set has not yet been compiled. The preliminary results do show, however, that the composite lifetimes are several times smaller than those predicted by the analytic Global Load Sharing model developed by the PI under the previous AFOSR grant. Also, the failure times are statistically distributed, as expected, a feature absent from any GLS model.

2h. Damage Evolution in Ceramic Matrix Composites

A new model for the evolution of matrix cracks in CMCs has been developed. The model assumes that matrix cracks emanate not from fiber-bridged flaws within the composite but rather from flaws in matrix-rich regions. These flaws then extend across the matrix-rich regions and propagate into the fiber-reinforced regions. Over a wide range of initial flaw sizes, the growth of the initial flaws into full-fledged matrix cracks extending across the entire composite is controlled by the size of the matrix-rich region rather than the initial flaw size. Thus, the number of matrix cracks versus stress, the spatial distribution of the matrix cracks, and ultimately the composite stress-strain curve are determined by the statistics of the matrix-rich regions of the composite and by the fiber/matrix interface sliding resistance that governs the fiber bridging once it becomes activated. Measurement of the matrix-rich regions coupled with the multiple-matrix-cracking model proposed earlier by the PI leads to predicted damage evolution in model glass matrix composites that is in good agreement with experiment.

2i. *Embrittlement of CMCs Due to High Interfacial Shear Stress*

We have investigated the possibility of decreasing ultimate tensile strength associated with increasing fiber/matrix interfacial sliding stress τ in ceramic matrix composites. There is some evidence that weakening, or "embrittlement" occurs for high τ yet no detailed model coupling realistic stress distributions and realistic flaw statistics to determine failure has been attempted previously. Here, we used a multiscale approach. An axisymmetric finite element model was employed to calculate axial fiber stresses versus radial position within the slipping region around an impinging matrix crack as a function of applied stress and interfacial sliding stress τ. The stress fields show an enhancement at the fiber surface, and we were able to develop a highly accurate yet simple empirical form for this stress concentration as a function of τ and the fiber radius. This stress field was then utilized as an effective applied field acting on annular flaws at the fiber surface, and fracture mechanics was used to calculate the mode I stress intensity as a function of applied stress, interface τ, and flaw size. A model for the probability of failure due to a pre-existing statistical distribution of flaws in the fiber was then developed. This was utilized within the Global Load Sharing model to predict fiber damage evolution and ultimate failure with the composite.

Our results show that for small fiber Weibull moduli ($m \approx 4$), the local stress enhancements are insufficient to preferentially drive failure near the matrix crack. Physically, it is more likely for larger flaws to fail at the lower stresses some distance from the matrix crack than for the high stresses near the crack to drive failure of smaller flaws. Hence, the composite tensile strength is weakly affected and follows the shear-lag model predictions, which show a monotonically increasing strength with increasing τ. On the other hand, for larger Weibull moduli ($m \approx 20$), the higher stresses near the matrix crack due drive preferential failure near the matrix crack. The composite is found to weaken beyond about $\tau = 50MPa$ and exhibit greatly reduced fiber pullout. Both of these features are consistent with an apparent embrittlement, and showing substantial differences as compared to the standard shear-lag model.

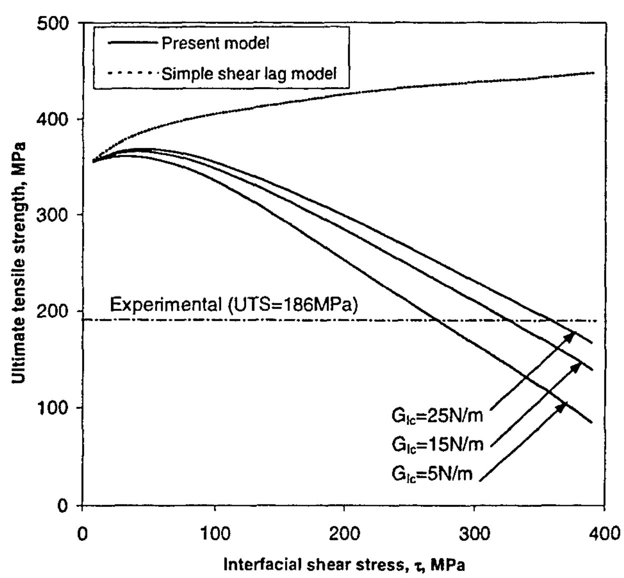


Figure 7. Tensile strength predictions (solid lines) versus interfacial shear stress τ for along with experimental strength. High τ can cause the low observed strengths.

Literature experimental data on Sigma SiC fibers with a high m in a glass matrix system was consistent with our predictions if τ is sufficiently large, although the value of τ for this system is not well-established (Figure 7). These results also suggest how embrittlement can occur due to interface oxidation, which can be interpreted in some cases as greatly increasing the interfacial τ.

2j. *Time-Dependent Failure of Ceramic Composites* (work primarily funded by NASA)

A micromechanically-based model for the deformation, strength, and stress-rupture life of a ceramic matrix composite was developed for materials systems that do not degrade by oxidative attack. The rupture model for a unidirectional composite incorporates fiber strength statistics, fiber degradation with time at temperature and load, the state of matrix damage, and the effects of fiber pullout, within a global load sharing model. The constituent material parameters required to predict the deformation and lifetime can all be obtained independently of stress-rupture testing through quasi-static tension tests and tests on the individual composite constituents. The model predicts the tertiary creep, remaining composite strength, and rupture life, all of which depend critically on the underlying fiber strength degradation. Sensitivity of the rupture life to various micromechanical parameters was studied parametrically. To complement the model, an extensive experimental study of stress-rupture in a Nextel 610/alumina-yttria composite at temperatures of 950°C, 1050°C, and 1100°C was undertaken. Constituent parameters for this system were derived from quasistatic tests and literature data, and the measured behavior were compared to the predictions of the model. The lifetime predictions were found to be two orders of magnitude smaller than observed. However, fitting the rate constant in the single fiber stress rupture behavior yields predicted lifetimes that agree will with the data over a range of applied stresses. The model then also predicts the tertiary creep and residual strength in good agreement with experiments with no further fitting. Reasons for the discrepancy in predicted and measured lifetimes may be due to aspects of the in-situ fiber strength degradation, and are discussed. The Larson-Miller and Monkman-Grant life prediction methods were shown to be inadequate for the current material.

2k. Stress Concentrations in Ceramic Matrix Composites

The "Global Load Sharing" model has been used almost exclusively in the modeling of CMCs. Yet, it is fully expected that localized stress concentrations exist that drive localized failure and these have never been investigated. Part of the difficulty lies in the fact that the problem requires a cracked matrix that regains load-carrying capacity away from the matrix crack plane. This precludes the use of shear-lag-type models commonly used in PMCs. To address this issue, we have used 3d FEM models to investigate the fiber-to-fiber stress concentrations in ceramic matrix composites as a function of various micromechanical parameters, particular interfacial sliding shear stress. We find that the stress concentration factors can be substantial, and that local load sharing is pertinent to CMCs. One outcome of local load sharing is a size dependence of the composite strength, and such a dependence has been demonstrated experimentally by other workers. We have not yet used the stress concentrations in our Green's function models to predict composite failure in a systematic manner, but have modified codes to do so for future workers who might be interested in pursuing this avenue of attack.

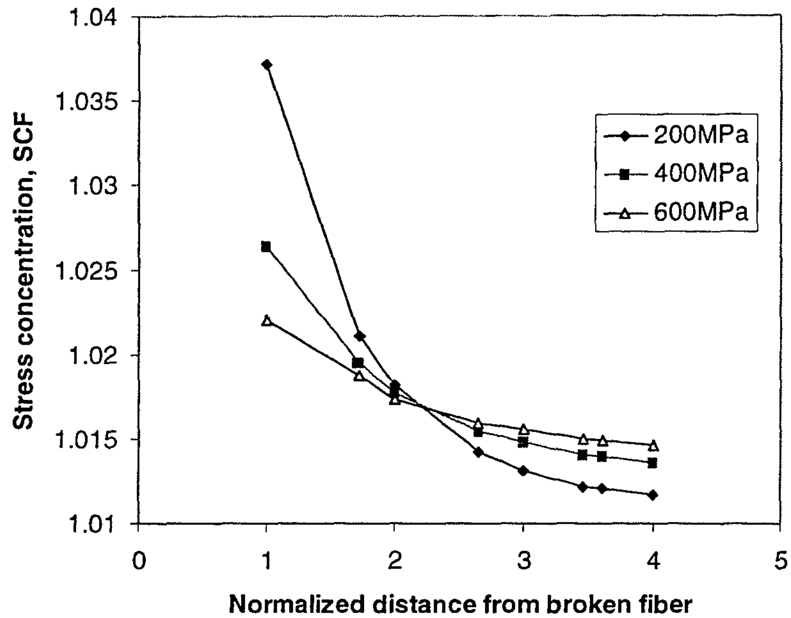


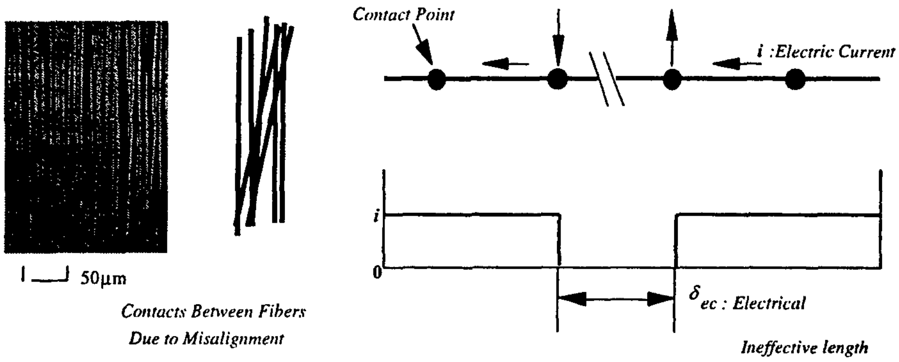
Figure 8. Stress concentrations around a broken fiber as a function of stress, in a SiC/SiC CMC with a single matrix crack

2l. Failure of Crossply Ceramic Matrix Composites
(work primarily funded by NASA)

The tensile failure of crossply ceramic matrix composites with a matrix through-crack has been examined numerically to assess the importance of stress concentrations at the 0/90 interface on failure. Fiber bridging in the cracked 0 ply was modeled using a line-spring model with evolving stochastic fiber damage introduced within the bridging model. A finite-element model was employed to determine the stresses throughout the cross-ply in the presence of the bridged crack. Results for two composite systems, SiC/SiC and a typical oxide/oxide, show that the stress concentrations at the 0/90 interface at low loads are largely relieved with increasing load due to the non-linear bridging response and preferential fiber damage near the interface. At the point of tensile failure, the region near the 0/90 interface is substantially damaged and the stress concentrations are essentially eliminated. Thus, the tensile strength is nearly identical to that of a unidirectional composite with the appropriate fiber volume fraction, a strength that can accurately be predicted analytically. Although this particular result was not unexpected, the presence of stress concentrations at lower loads will preferentially drive stress rupture at high temperatures and lead to reduced composite lifetimes. This model has been adapted, using our prior analytic models for unidirectionals, to predict the time-dependent stress rupture behavior. Results indicate that crossply composite lifetimes are reduced relative to unidirectional composites by factors of 2-10, depending on the micromechanical parameters relevant to the fiber strength degradation.

2m. Electrical Resistance as a Sensor of Damage in CFRP: a GLS Model

Having successfully modeled mechanical damage evolution and failure in various systems using multiscale models in new ways, we have moved forward into the area of damage detection in graphite-fiber/epoxy composites. The structural carbon fibers in a polymer-matrix/carbon-fiber composite are electrically conducting while the polymer matrices are electrical insulators. Hence, a composite ply consisting of more-or-less well-aligned carbon fibers exhibits a finite electrical resistance associated with the resistance of the carbon fibers. Under increasing applied load, the carbon fibers break at the locations of flaws in the fibers. Each fiber break interrupts the electric current flow through the broken fiber and increases the longitudinal (along the fiber direction) resistance of the composite, so that resistance is somehow a measure of mechanical damage. Our first work in this area has focused on global mechanical and electrical models that relate damage to resistance change in a quantitative manner.



(a) (b)

Figure 9. a) Contacts between fibers; b) schematic of "electrical ineffective length" concept.

Fibers are not perfectly aligned and the intermittent contacts between fibers form additional pathways for the current to flow. Due to these fiber contacts, current flows in the broken fiber up to those contacts nearest to the break and is then diverted around the fiber into the neighboring contacting fibers. Only a portion of the fiber length no longer participates in the current-carrying capability of the electrical network. In other words, around each fiber break there is a zone, the "electrical ineffective length" over which the fiber cannot carry any current.

The "electrical ineffective length" is analogous to the "mechanical ineffective length" δ around the break: further than $\pm\delta$ from a break location, the fiber carries a full load and the composite appears undamaged. There is, furthermore, a characteristic mechanical ineffective length δ_c that is a combination of mechanical properties and statistical fiber properties and that is the relevant length for describing damage evolution in the composite. We have proposed the existence of a "characteristic electrical ineffective length" δ_{ce}, which is nominally the typical distance between electrical contacts of nearby fibers in the composite. Using this new idea, and our early models for stochastic fiber damage up to failure, a prediction for the electrical resistance as a function of strain (upon changing from stress to strain using the fiber Young's modulus), and including the well-known linear dependence of the single-fiber resistance on strain, can be expressed as

$$\frac{\Delta R}{R_o} = (1+\alpha\varepsilon)\, e^{\frac{\delta_{ec}}{\delta_c}\left(\frac{\varepsilon E_f}{\sigma_c}\right)^m} - 1 \qquad (1)$$

where α is the known linear strain/resistance response coefficient for the single fibers and σ_c is the characteristic fiber strength identified originally by the PI. The predictions of Eq. 1 are shown in Figure 11, and the agreement with just one fitting parameter (all of the fiber statistical strength quantities, modulus, and α are independently measured) is quite remarkable. The independence of the electrical resistance to the sample gauge length predicted by Eq. 1 is also validated by experiments, as shown in Figure 10.

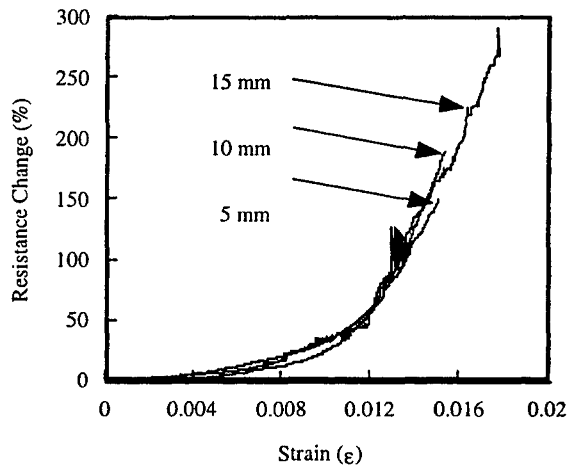


Figure 10. Normalized resistance change versus strain as measured and as predicted by Eq. 1 for various values of the "electrical ineffective length"; the value $\delta_{ec} = 5mm$ fits the experimental data over the entire strain range.

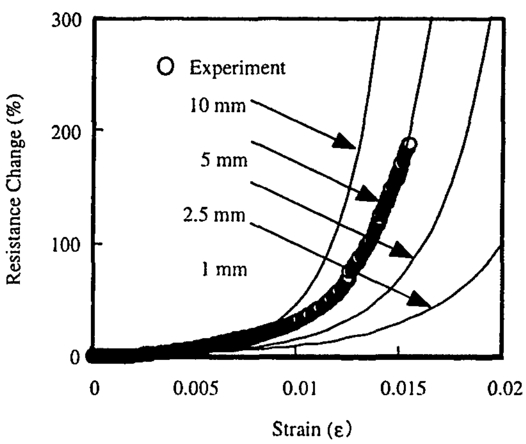


Figure 11. Normalized resistance change versus applied strain for samples of different gauge length.

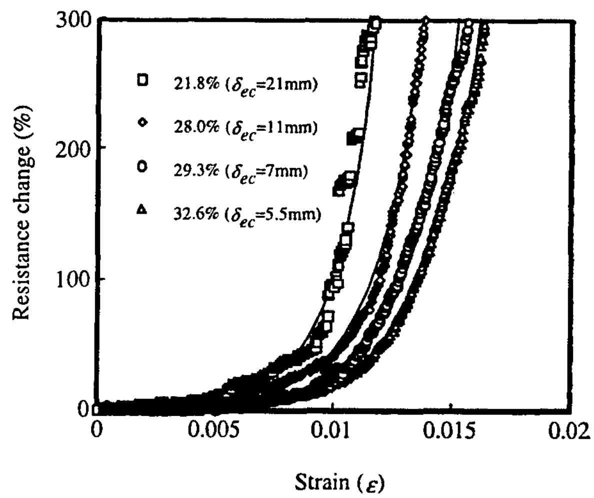


Figure 12. Normalized resistance change versus applied strain for samples of different fiber volume fraction. Solid lines show predictions from Eq. 1.

The degree of fiber/fiber contacts and the associated length δ_{ce} are geometrical quantities. Takeda and coworkers have recently shown that the resistance versus strain depends on the fiber volume fraction, and the model of Eq. 1 can predict the measured resistance with a volume-fraction-dependent δ_{ce}, as shown in Figure 12. The dependence of δ_{ce} on volume fraction has been calculated using a 2-dimensional percolation analysis for misaligned rigid rods, and reasonable correlations with the values indicated in Figure 12 have been obtained.

2n. Electrical Resistance as a Sensor of Damage in CFRP: Coupled Numerical Model

The success of the global model above was exciting, but such models have distinct limitations. In particular, (i) they are not sensitive to the localized damage that is actually driving failure, (ii) there is no direct verifiable coupling of the electrical ineffective length to actual fiber/fiber contacts, and (iii) there is no dependence on the geometry of the voltage leads. The mechanical "global load sharing" models have been known to overestimate strength in PMC systems, and hence are not currently widely used; so-called Local Load Sharing models are applied. Our work here as been to develop a coupled electrical/mechanical model that accounts explicitly for the effects of individual fiber breaks on both mechanical and electrical response, holds the capability for local detection of damage, and can show the dependence of the electrical response on voltage lead geometry. This preliminary work will lead to future work that can guide the detailed development of self-sensing CFRP systems.

To quantify the dependence of electrical resistance on mechanical fiber damage at the micromechanical level, a numerical electrical resistor network model has been developed. The electrical model is coupled to the fiber damage through a numerical mechanical model, which here is a shear-lag model; the latter informs the electrical model of the locations of broken fibers and the stresses on unbroken sections of fiber. Figure 13 shows a schematic of the development and coupling of the two models. The electrical model accounts for both the longitudinal fiber electrical conductivity and the occasional fiber-fiber contacts that permit electrical coupling of touching fibers.

The analytic model based on the Global Load Sharing theory has been tested against the numerical simulations, and good agreement is found for voltage leads that contact the ends of all fibers in the composite and when the "electrical ineffective length" in the GLS model is related to the density of fiber contacts f_c in the numerical model by the relationship $\delta_{ec} = L/(1+f_c L)$ where L is the sample gauge length, as shown in Figure 14.

Voltage leads that contact only surface fibers lead to a resistance behavior that cannot be explained by the GLS model, and demonstrate the spatial sensitivity of the electrical response to damage and the necessity of simulation models, as shown in Figure 15. Sensitivity of the response to the voltage lead geometry suggests that the coupled numerical model can be used to design electrode arrays to optimize damage detection inside composite structures. Further investigations of the spatial sensitivity of damage detection versus the fiber contact density remains to be carried out.

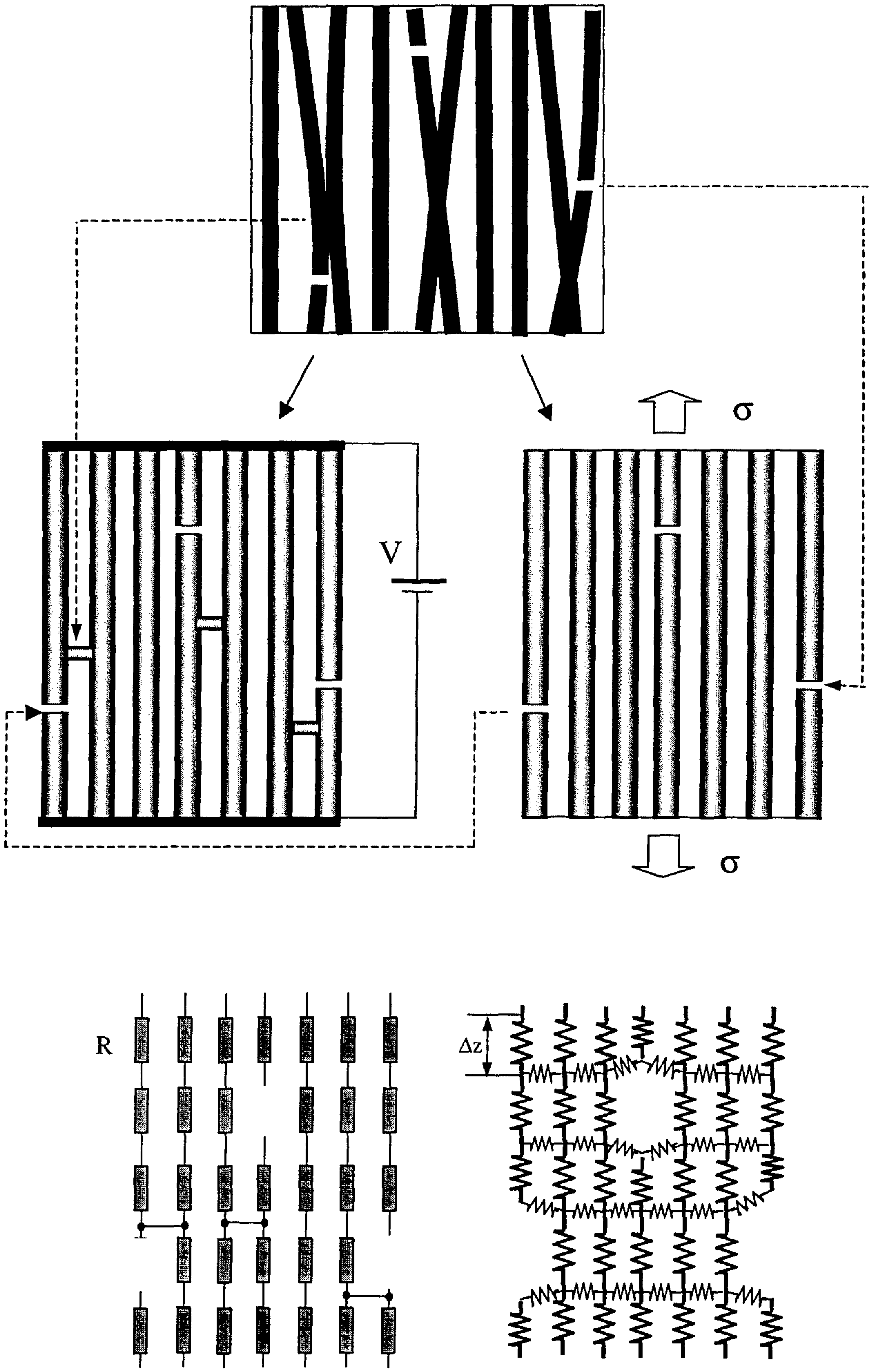


Figure 13. Schematic of coupled electromechanical model.

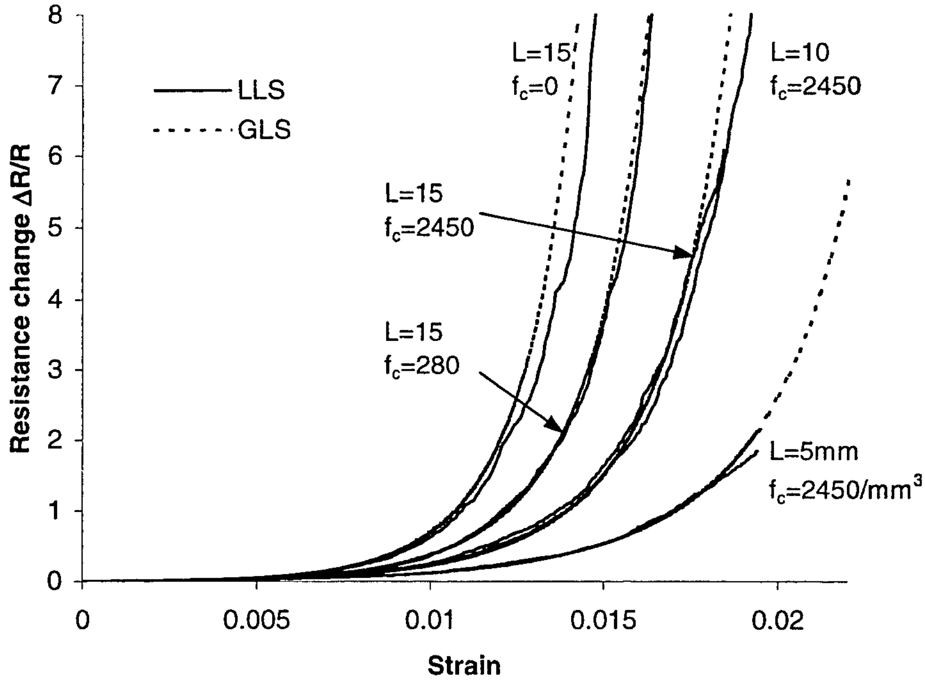


Figure 14. Electrical resistance versus strain, as predicted by the computational LLS model and analytical GLS model for various fiber contact fractions and for end voltage lead contacts.

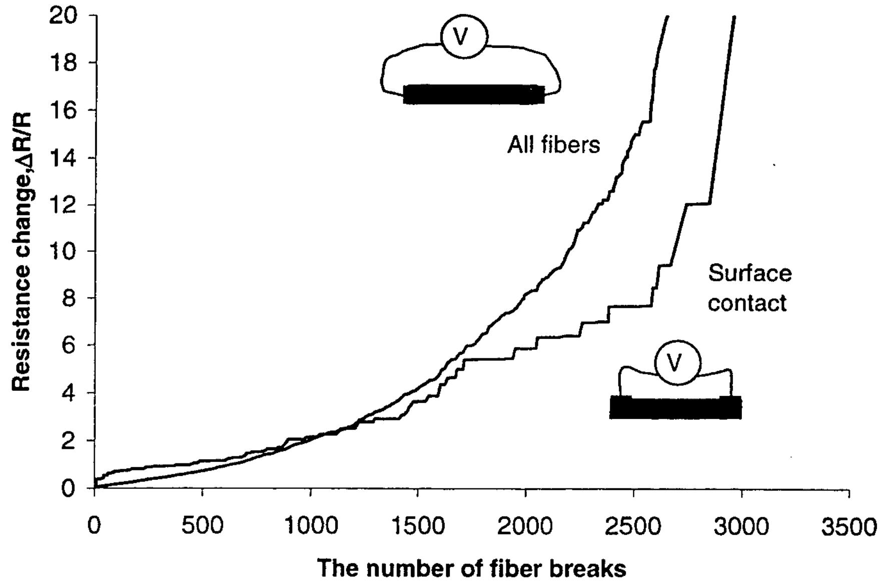


Figure 15. Electrical resistance versus number of fiber breaks for both end and surface voltage lead contacts. Note the difference in response simply due to voltage lead geometry.

3. Publications generated under grant funding

1. W. A. Curtin, “Stochastic Damage Evolution and Failure in Fiber-reinforced Composites”, Advances in Applied Mechanics 36, 163-253 (1998). **(invited)**

2. W. A. Curtin, "Stress-strain response of Brittle Matrix Composites", Chapter 3 in Encyclopedia of Composites Vol. 4, Eds. A. Kelly, 2000, pp. 47-76. **(invited)**

3. W. A. Curtin, “Dimensionality and Size Effects on the Strength of Fiber-reinforced Composites”, Comp. Sci. Tech. 60, 543-551 (1999).

4. W. A. Curtin, “Tensile Strength of Fiber-reinforced Composites III. Beyond the Traditional Weibull Model for Fiber Strengths”, J. Comp. Matls. 34, 1301-1332 (2000).

5. T. Okabe, N. Takeda, J. Kotomori, M. Shimizu, and W. A. Curtin, “A New Fracture Mechanics Model for Multiple Matrix Cracks of SiC Fiber Reinforced Brittle Matrix Composites”, Acta Mater. 47, 4299-4309 (1999).

6. Z. H. Xia, W. A. Curtin, and P. Pieters, “Multiscale Modeling of Failure in Metal Matrix Composites”, Acta Mater. 49, 273 (2001).

7. Z. H. Xia and W. A. Curtin, “Tough to Brittle Transitions in Ceramic Matrix Composites with Increasing Interfacial Shear Stress”, Acta Mater. 48, 4879 (2000).

8. Z. H. Xia and W. A. Curtin, “Life Prediction of Titanium MMCs Under Low-Cycle Fatigue”, Acta Mater. 49, 1633 (2001).

9. Z. H. Xia and W. A. Curtin, “Multiscale Modeling of Failure in Aluminum Matrix Composites”, Comp. Sci. Tech. 61, 2247 (2001).

10. T. Okabe, N. Takeda, Y. Kamoshida, M. Shimizu, and W. A. Curtin, “3D Shear-lag Model Considering Microdamage and Statistical Strength Prediction of Unidirectional Fiber Reinforced Composites”, Comp. Sci. Tech. 61, 1773 (2001).

11. Z. Xia, T. Okabe, and W. A. Curtin, “Shear-lag Versus Finite-element Models of Deformation in fiber composites”, Comp. Sci. Tech. 62, 1141 (2002).

12. Z. Xia, T. Okabe, and W. A. Curtin, “Green’s Function Versus Shear-lag Models for Damage Evolution and Strength Predictions in Fiber Composites”, to appear in Comp. Sci. Tech. (2002).

13. H. Halverson and W. A. Curtin, “Stress-rupture in Ceramic Matrix Composites: Theory and Experiment”, J. Am. Cer. Soc. 85, 1350 (2002) (**Feature Article**)

14. J. B. Park, T. Okabe, N. Takeda, and W. A. Curtin, "Electromechanical Study of the Internal Conducting Network of CFRP Composites", Comp. Sci. Tech. (2001).

15. J. B. Park, T. Okabe, N. Takeda, and W. A. Curtin, "The Electromechanical Modeling of Unidirectional CFRP Composites Under Tensile Loading Conditions", Comp. Part A. 33, 267 (2002).

16. T. Okabe, K. Misawa, M. Yanaka, M. Shimizu, N. Takeda, and W. A. Curtin, "Statistical Studies on Cracking Phenomena in TiN Thin Film on Titanium Substrate", submitted to Acta Mater. (2001).

17. M. O'day and W. A. Curtin, "Failure in Crossply Ceramic Matrix Composites", J. Am. Cer. Soc. 85, 1553 (2002).

18. Z. Xia, T. Okabe, J. B. Park, W. A. Curtin, and N. Takeda, "Damage Detection in CFRP Composites: Coupled Mechanical and Electrical Models", subm. Acta Mater. (2002).

4. Presentations related to work under this grant (invited only)

"Damage Evolution and Failure in Heterogeneous Materials", Yale U., 2/99.

"Time-dependent failure of oxide/oxide composites", NASA Glenn Res. Ctr., 4/99.

"Damage Evolution and Failure in Heterogeneous Materials", Aerospace Dept., U. Minn, 5/99.

"Mechanical Behavior of Ceramic Matrix Composites", DLR Institute, Cologne, Germany, 8/99.

"Damage Evolution and Failure in Heterogeneous Materials", DLR Institute, Cologne, Germany, 8/99.

"Damage Evolution and Failure in Heterogeneous Materials", U. Conn., 3/00.

"Time-dependent failure of oxide/oxide composites", NASA Glenn, 4/00.

"Damage Evolution and Failure in Heterogeneous Materials", Ohio St. U., 4/00.

"Heterogeneity in Composite Materials: Good or Bad?", ARO Workshop on Heterogeneous Materials, 6/00.

"Multiscale Models of Failure in Fiber-Reinforced Composites", Intl. Congress

Theoretical and Applied Mechanics, Chicago, 8/00.

"Multiscale Modeling of Failure in Fiber Composites", Workshop on Advances in Continuum Damage Mechanics, Cachan, FR, 9/00

"Damage Accumulation and Failure in Fiber-Reinforced Composites", (**4 lectures**) Ameland Summer School on Micromechanics, (Delft University, Holland), 10/00.

"Coupling Atomistic and Continuum Models of Deformation", ARO Workshop on Multiscale Methods (Aberdeen, MD), 5/01.

"Coupling Atomistic and Continuum Models of Deformation", CECAM Workshop on Multiscale Methods (Crete, Greece), 7/01.

5. Personnel supported on this grant

W. A. Curtin, Professor of Engineering and PI

Dr. Zhenhai Xia, Research Associate

Mr. Michael P. O'day, Graduate Student in Solid Mechanics